NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

TECHNICAL MEMORANDUM

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THE FURTHER DEVELOPMENT OF HEAT-RESISTANT MATERIALS

FOR AIRCRAFT ENGINES

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The present report deals with the problems involved in the greater utilization and development of aircraft engine materials, and specifically piston materials, cylinder heads, exhaust valves, and exhaust gas turbine blading. The blades of the exhaust gas turbine are likely to be the highest stressed components of modern power plants from a thermal-mechanical and chemical standpoint, even though the requirements on exhaust valves of engines with gasoline injection are, in general, no less stringent. For the fire plate in Diesel engines the specifications for mechanical strength and design are not so stringent, and the question of heat resistance, which under these circumstances is easier obtainable, predominates.

The development of materials for exhaust gas turbines will primarily be based on the proved, well-known and still suitable exhaust gas materials, several of which are given in table 1.

TABLE 1.- COMPOSITION OF VARIOUS EXHAUST VALVE MATERIALS

Number	Alloying parts in percent							
	C	Si	Mn	Ni	Cr	W	Mo	Fe
1	0.45	1.0	1.0	14	12	2.0	--	Remainder
2	.50	2.0	.4	12	22	3.0	--	Do.
3	.45	1.6	1.4	28	15	4.0	--	Do.
4	.02	1.0	1.9	60	15	.40	7.0	14

[1]"Über die Weiterentwicklung warmfester Werkstoffe für Flugzeugtriebwerke." Luftfahrtforschung, vol. 14, no. 4/5, April 20, 1937, pp. 196-203.

The high temperatures encountered in these valves cancel martensitic steels which, although highly resistant to scaling (such as Cr-Si-steel), show considerably lower resistance to elevated temperatures and lower creep strength above 500° to 550° C. Higher temperatures result in very high thermal stresses because of possibly greater temperature gradients. In addition, there are the purely mechanical stresses due to external and mass acceleration forces. In most martensitic steels the critical temperatures lie below the maximum temperatures encountered in exhaust valves. The temperatures that lead to form and structure changes without transformation (tempering effect: balling and redissolution of carbide) as well as to recovery recrystallization and crystal growth, are too low. Aside from the strength requirement at elevated temperatures, the necessity for stability of the chemical and mechanical properties also stipulates the use of alloys free from transformations (reference 1), which, for steel, would suggest the use of **austenitic** alloys with close-packed atomic arrangement. Precipitation and the agglomeration of carbide precipitates and intermetallic compounds, especially of the carbides in austenitic steels containing carbon, are successfully prevented by special alloying elements, or at least slowed down considerably by dispersion so as to avoid the danger of intergranular corrosion and embrittlement. Titanium, tantalum, and columbium are the principal metals for stable carbides. Although even here the stability is not perfect, it is sufficient for practical purposes.

Creep and fatigue strength at high temperatures are more important properties for exhaust gas turbine blades than for exhaust valves. Whereas the short periods of high stress in the exhaust valves are followed by comparatively long periods of almost complete relaxation, the blades and impellers are under continuous maximum stresses throughout the entire period of operation.

Of great importance are eventual superimposed alternating stresses due to either periodic impulses in partially impacted turbines, or the very high natural vibration frequencies of the relatively short blades compared to steam turbines. At the present state of development of the exhaust gas turbine the experimental data on blade vibrations and induced alternating stresses are far from complete. It will be necessary however, to analyze, aside from the mean stresses and local stress increments due to the design, the possibility of the existence of vibrations particularly. Figure 1 illustrates failures of superheated steam turbine blades which failed at stresses below the theoretical creep strength, probably as the result of vibrations. Whereas the corrosion stability in exhaust valves is already an indispensable property, this holds even more so for the efficiency and safety of turbine blades. What relationship the stability against corrosion and erosion in the high-speed exhaust gas bears with those found in scaling apparatus and on exhaust valves has not yet been investigated.

On reverting to the valve steels as a basis for the present arguments, the creep strength is of primary importance. Figure 2 gives the creep strength values at elevated temperatures of several commercially known heat- and scale-resistant alloys. In view of the exhaust gas temperatures of airplane engines (reference 2) - less in Diesel than gasoline engines - according to table 2, and the peripheral speeds for the blade rims required for satisfactory efficiency, these creep strength figures cannot be exactly regarded as high or about sufficient. Therefore, the use of exhaust gas coolers at least between internal combustion engine and exhaust gas turbine seems indispensable. But even if a drop in exhaust gas temperature of, say to 700° or rather even to 650° to about 580° C up to entry in the impellers is assumed, the creep strength figures for the still necessary peripheral speeds of from 320 to 370 meters per second are not sufficient. The tensile stresses to be expected in exhaust gas turbines at exhaust gas temperatures of around 580° to 600° C and for an output of 100 to 130 horsepower at the blade root, lie between 20 and 25 kilograms per square millimeter. The stresses, almost exclusively centrifugal are proportional to the specific weight of the material, so that the specific weight is also a factor to be considered in the development of the materials. Considering that for the operating temperatures aimed at, short-period temperature rises of 600° to 650° C must be taken into the bargain under certain circumstances, the unfavorable material situation is readily apparent.

If a less favorable utilization of the exhaust gas energy through further intercooling or lowered turbine efficiency is to be avoided, two avenues separately or conjointly are open: First, improving the strength at elevated temperatures and second, the admission of stresses above the creep strength. The best commercial scale-resistant and heat-resistant materials for exhaust valves of today were developed in the most essential properties as long as 10 or more years ago. Up to now the structure stability and corrosion strength have been improved, but as far as the afore-mentioned kinds of alloys are concerned, the development appears to have nearly reached the best obtainable values.

There are only three possibilities to manufacture materials with greater strength at elevated temperatures and adequate resistance to scaling: First, by favorable composition with greatest possible stability of the constitution of the structure; second, through relaxation and recrystallization temperatures that lie sufficiently above the operating temperatures, strain-hardening by cold working; and third, precipitation hardening for operating temperatures that lie so far below the precipitation temperatures that even operating stresses induce no continuous precipitation and, above all, no agglomeration of the precipitation.

An analysis of the effects of all alloying elements involved here would lead too far beyond the scope of the present report. Therefore the results of various tentative experiments made in the sketched direction are discussed.

In precipitation-hardenable austenitic alloys the precipitation is slower and occurs at higher temperatures than in ferrite alloys. Accordingly, the coagulation of the precipitation is slower. Now, there are a number of known alloys with very high precipitation temperatures between 600° and 900° C, which change their characteristics very gradually even for very long annealing temperatures after reaching maximum values - for instance, Co-W-Fe, Co-W-Mo, Fe-C-Cr-Ni-Be, Fe-C-Ni-Cr-Ti, furthermore Fe-Cr-Ni-alloys with addition of B and Al. Of these several Co-W-Fe-alloys with the compositions as given in table 2 were analyzed. Several of the test data have been reported elsewhere (reference 3). In all these cases it involved cast, nonmalleable samples of great brittleness, the use of which does not appear advisable for exhaust gas turbines owing to the low notched impact values and danger of failure even with the least notches. The precipitation hardening at 600° to 750° C was unusual, for instance, for melt 1 to 3 from $R_c = 30$ to 34 to $R = 50$ to 65, for melt 5 from $R_c = 2$ to $R_c = 60$. At annealing temperatures up to 700°, annealing periods of up to 300 hours disclosed but a trifling reduction in hardness. But the creep strength at 600° and 700° was below that of the austenitic Cr-Ni steels, according to figure 3. More searching investigations of the changes in structure of the supercooled alloys for annealing disclosed the existence of a cubic face-centered (austenite) lattice in the ground mass in the supercooled state, but contrary to conceptions held heretofore, the austenitic lattice at temperatures below 800° appeared unstable; hence was not preserved in the hardening.

The ground mass of melt Nos. 1 to 3 supersaturated with $CoW - Fe_3W_2$ rather transformed itself in an intermediary type of crystal. The supercooled alloys 4 and 5 assume a hexagonal lattice structure on annealing (β-γ conversion of cobalt) and precipitate an intermediary crystal type. In this temper-hardening there is the danger that the alloys are subjected to corrosion attack, particularly intergranular, when the precipitations occur chiefly on the grain boundaries or on the slip bands at deformation. A decrease in notched-bar impact strength is usually associated with these forms of precipitation.

According to this, the probability of securing adequate strength by precipitation hardening at elevated temperatures for the operating temperatures of around 600° to 650° C is not very great. However, the research has just started, so that no definite judgment can be given as yet.

With regard to increasing the creep resistance and the flow resistance through cold working, the relaxation from the consequence of cold working was analyzed on several austenitic steels. The results are published elsewhere (reference 4). The steels involved are given in table 4.

In tests on the effect on the hardness and notched impact strength in slow annealing at 600° to 900° of quenched and cold-worked steels Nos. 1, 2 and 3; steel No. 3 canceled out for further study because of its marked embrittlement. Since steel No. 1 tends toward intergranular corrosion at elevated temperatures in consequence of carbide precipitation at the grain boundaries, steel No. 2 was selected for the study of the effect of cold working on the creep resistance. But the composition of this 15 to 20 percent cold-worked bar (steel No. 4) differed from that of steel No. 2, especially in Ti content. The inferior behavior of the cold-worked compared to the quenched steel at 600° to 700° C is attributable to the greater plasticity associated with the appearance of the recovery of the strength characteristics. Whereas steel No. 2 exhibited no perceptible recovery of hardness and notched-impact strength at the foregoing annealing tests at 700° C, steel No. 4 manifested a partial relaxation at 600°, according to the static endurance tests. Hardness measurements on cold-worked specimens of steel No. 4, when exposed without outside stress and under constant tensile stress at temperatures of 500°, 600° and 700°, revealed that the recovery of the strength characteristics through outside tensile stresses is shifted toward lower temperatures, and is more complete.

Greater strength at elevated temperatures is therefore most likely to be attained by suitable manufacture and working on the basis of the chemical composition. Several experiments in this direction will be discussed.

Earlier in the report the suggestion for better utilization of the materials by increasing the stress beyond the creep strength, as outlined by the sub-committee for tensile testing of the Society of German Metallurgists (reference 5), had been made. That this can be accomplished only with sufficiently safe creep behavior under the operating stresses for at least the operating period, requires no special demonstration. Since the cited directives provide an abbreviated method for temperatures up to 500° to 550° C at the most, it must be modified for higher test temperatures. For the transition from an originally higher rate of stress to constant or static takes from 150 to 250 hours at elevated temperatures. The life of airplane exhaust gas turbine blades compared to stationary steam turbines must, for a multiplicity of reasons, be chosen proportionally short. With the valve steels available at present a life expectancy of between 250 and 500 hours is justifiable. Therefore, the tests were extended above the time-yield curves on an average of over 300 hours and were carried out in air furnaces. Several of the test data are given in table 5. The alloys Nos. 4, 7, and 9 were melted in the high-frequency furnace of the Institute for Material Research of the DVL. Figure 5 shows the rate of creep of several of these alloys at and above the creep strength and for temperatures of 600°, 650°, and 700° C. After exceeding the load at which the rate of elongation ultimately is equal to zero, two cases can occur: The first decreasing rate of creep is followed by a constant creep rate which ultimately results in failure, or else the high creep rate at the start of the time load decreases temporarily, increases again and then also leads to failure but within a shorter time. In this instance a well-known phenomenon is very unfavorable: The great elongations observed in the high-temperature tensile test cannot be utilized by far and, accordingly, present no high safety factor against failure. Thus, for example, alloy No. 8 has an elongation at rupture of 43 percent in the tensile test at elevated temperature of 600°, as against the 15 percent elongation in the creep test at the same temperature. The differences are very much greater for other alloys. Alloy No. 3 indicated under 21 kilograms per square millimeter load and 650° a creep elongation of 1 percent up to failure, and a creep elongation of less than 4 percent for 18 kilograms per square millimeter and 700° C, while the tensile test at 600° C indicated an elongation of 20 percent. The creep was accompanied by practically no constriction; the break occurred almost without deformation. The tests must be extended so far that failure at the chosen load and duration, and especially any secondly increasing rate of creep, is safely ruled out. Particularly promising is the DVL alloy No. 7, according to figure 5. It has the advantage that the creep resistance with increasing temperature to 650° and 700° C shows no expressly steep drop. This ensures greater safety at short-period elevated temperatures. The same applies to the DVL alloy No. 4, which, it is true, is not situated as well as Nos. 1, 2, 3, and 7 at low temperatures of 600° C. For better comparison the total elongations and creep rates from the creep tests are plotted for several test temperatures (figs. 6, 7, and 8). The possibilities discussed earlier for the design are indicated by shaded areas. Permissible total elongations of 2 and 3 percent in 300 hours and loads from 20 to 22 kilograms per square millimeter served as basis for the material selection in figures 6 and 7. The best of all is alloy No. 7; then follows Nos. 3, 2, 6 (No. 6 is rejected because of unsatisfactory corrosion resistance), 1, 9, 4, and 8. At 700° C the order is Nos. 7, 4, 5, and 2 so far as the tests had been made.

Alloys with satisfactory strength at elevated temperature and high creep rate often show a comparatively high initial elongation. It might be possible to take away the initial elongation from the operation beforehand by pre-loading at operating temperatures. In this event rate of creep is the chief factor, according to figure 9. The creep rates leading to creep elongation of from 2 to 3 percent in 700 hours are emphasized. The evaluation according to creep rate, however, leads to the same order as that based on the total elongation. The final amounts of the extreme elongations of from 2 and 3 percent are, of course, not reached then.

The effect of heating under load, relaxation and reloading at elevated temperatures was also studied in individual cases. This, as well as the effect of nonuniform heating occurring at starting and stopping of engines with high output, will be summarized later. In this case a high local and momentary heat gradient will appear so much more as the blade sections are more tapered. The materials must at these stresses yield high plastic elongations and contractions. This also is the case for blades cooled inside or at the root. A measure for practical use is the number of temperature reversals until permanent deformations and finally cracks occur. Several examples of this test series are reproduced in figure 9. The highest temperatures were, of course, considerably above those considered reliable in operation.

Even less explored are the blade vibrations and vibratory stresses in blades, superimposed on the elastic stresses.

And of the effect of alternating stresses and the alternating stress superposed on a constant mean stress on the strength and strains at elevated temperatures, at least as far as austenitic alloys are concerned, almost nothing is known. Various data published in the literature are reproduced in table 6. The values for the resistance to alternating stresses (fatigue strength) were obtained by a method patterned after the Wöhler test. According to the latest investigations (reference 6) the stress distribution under tensile-threshold load at elevated temperatures is nearly the same as in creep tests in which the load is equal to the highest load of the tensile-threshold load. Therefore the Wöhler curve is impractical as basis for the design. It rather seems that the creep test could replace the fatigue test.

Like the valve steels, the light alloys for pistons and cylinder heads have reached their present state of development for many years. Table 7 correlates several of the most important and widely used Al piston alloys. The requirements on piston alloys are high and varied. The controlling factors are great strength at elevated temperature, low heat expansion, high thermal conductance, and good running characteristics. The temperatures in the bottom of the piston go to beyond 300° C and in the stem up to 180° to 200° C. Up to now tests on the fatigue strength at elevated temperatures were lacking. The investigations of the Institute for Material Research yielded table 8. Naturally, the same statement made previously for the fatigue stress of the steels resistant to elevated temperatures holds true also for these alternating bending tests. Since in the bending on the rotating bar any plastic deformation is invariably made retrogressive, such tests afford no information about the behavior under expanding load as actually occurs in piston and cylinder head. The most important and still necessary tests in this direction are creep tests. Even if at start of operation the alloys originally have sufficient mechanical strength, it nevertheless should be borne in mind that these characteristics are subject to considerable changes during operating periods of 200 hours. According to investigations in the Institute for Material Research, decreases, as indicated in figure 12 and table 9, were obtained.

According to it and structure studies these alloys are not invariable relative to protracted annealing at operating temperatures, and specifically, the forged alloys are a little more susceptible than the cast alloys. However, forgings are superior to castings as far as strength is concerned, because their more favorable fiber arrangement makes for greater homogeneity and fewer defective spots. This explains the preference accorded to forgings over castings, although the latter, especially those with high silicon content, show less wear and better gliding characteristics.

Casting and quenching stresses, likewise structure variations at annealing, produce deformations (reference 7) which can be prevented and made harmless by heat treatment before finishing and installing.

But in the latest power plants the conventional alloys are scarcely sufficient. For particularly unfavorable conditions in Diesel engines the piston types have been successfully protected from the fine-pointed flame by "fire plates" of highly heat-resistant steel.

Studies carried on within the last years conjointly with the Institute for Electrometallurgy and Metallurgy of the Technical High School at Aachen have led to the development of Al alloys with greater strength at elevated temperatures and high wear resistance. The aim was to combine the advantage of a low heat expansion of the alloys with high Si content (such as alloy Nos. 4 and 6) with the superior mechanical characteristics of an alloy such as Nos. 1 and 4 into a forgeable alloy. The goal was achieved in a rather complex alloy consisting of: Si 10.5 percent, Fe 1.5 percent, Cu 1.5 percent, Mg 0.9 percent, Co + Ni + Cr + Ti + Ag = 3.8 percent, and the rest Al. The Ag content is for good forgeability.

The forged alloy is age-hardenable and shows a strength increase from 90 to 100 up to 155 to 165 kilograms per square millimeter (Brinell) after quenching at 490° to 500° C in water and annealing at 180° to 200° C for 20 hours. The slow decrease in hot-hardness with rising temperature, compared with alloy No. 4 (table 7) in figure 13, is of advantage. The structure is unusually fine-grained in the casting state, according to figure 14. The microstructure of this alloy in the forged and age-hardened state is shown in figure 15. The very fine distribution of all alloying constituents, especially of the heavy metals, is obtained by appropriate melting process in the form required for high wear resistance and for good running characteristics.

The evaluation of the Al alloys for cylinder heads is based on the same arguments as for the piston alloys but minus the wear resistance and running characteristics. Here also the forged cylinder head has many distinct advantages over the cast head. Strength at elevated temperature and freedom from defects places the alloys Y, RR 53 and RR 59 at the head of the list.

From a survey of the findings, the inescapable conclusion seems to be that for many materials no important improvement in properties may be anticipated right away. But there is a promise that further development of isolated essential properties will secure greater uniformity and hence dependability. A number of proved alloying constituents are indispensable according to the present stage of research.

Translation by J. Vanier,
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TABLE 2

DISCHARGE TEMPERATURE OF EXHAUST GASES FROM AIRCRAFT ENGINES

Type of power plant	Exhaust gas temperature at exit from cylinder	
Diesel engine	450-500-700	Depending upon type and load
Gasoline engine - no supercharger (carburetor engine)	950-1150	
Gasoline engine supercharger (carburetor engine or injection engine)	1000-1200	

Table 3.- Composition of Co-W-Fe alloys.

No.	Alloying constituents in %						
	C	Mn	Si	Co	W	Al	Fe
1	0,02	0,03	0,07	38,9	40,7	0,39	Remainder
2	0,02	0,06	0,05	38,2	39,4	0,4	"
3	0,05	—	—	40,1	36,0	Trace	"
4	—	—	—	69,4	20,4	—	"
5	—	—	—	55,8	25,2	—	"

Table 4.- Chemical composition and heat treatment of steels after cold-working.

No.	C	Composition in %						Quench temp.	Structure
		Si	Mn	Cr	Ni	Mo	Ti		
1	0,12	0,73	0,63	17,24	9,37	0,22	Trace	1130° C	γ
2	0,12	1,16	0,61	18,00	9,58	0,10	0,42	1070° C	γ
3	0,12	1,16	0,62	18,37	9,03	2,40	0,36	1080° C	$\alpha + \gamma$
4	0,10	0,6	0,33	18,3	9,0	0,24	0,17	1080° C	γ

Table 5.- Chemical composition of various alloys used for the Rate of Creep tests - test data -

(Duration of test 300 hours)

No.	Composition in %						Rest	Test temp. °C	Stress kg/mm²	Total elong. %	Mean rate of creep-%/hour
	C	Cr	Ni	W	Mn	Si					
1	0,5	13—15	13—15	2,5—3,5	1	0,8	Fe Rest	590	21,5	0,82	0,0023
								600	18,1	0,93	0,0029
									21,5	1,055	—
									24,5	1,23	0,0033
								650	11,9	0,29	0,0007
									15,2	1,11	0,0033
2	1,0	13—15	13—15	2,5—3,5	1	0,8	Fe Rest	600	15,1	0,25	0,0005
									19,8	0,42	0,0008
									25,3	0,455	0,0028
								650	11,9	0,55	0,0015
									15,2	1,20	0,0034
									18,2	4,59	0,0149
								700	8,15	0,35	0,0010
									12,5	3,3	0,0107
3	0,1	19	10	0,5 or 0,2 Mo	0,7	0,3	1,3 Ta + Nb, Rest Fe	600	18	0,151	0,00007
									22,35	0,524	0,0008
									24,5	0,43	0,00137
4	0,25	15	20	10	1,5	3,0	Rest Fe	700	5,4	0,05	0,00007
									10,3	0,38	0,00097
									15,4	1,25	0,0037
5	0,31	13	10	—	1,0	1,58	Mo 7,09 Rest Fe	700	5,4	0,22	0,00067
									7,45	0,20	0,00050
									9,65	0,77	0,0019
									15,1	2,00	0,0043
								800	1,9	0,39	0,00127
6	0,25	12,3	2,2	—	12,6	0.38	8,9 Mo	600	19,5	0,5	0,00127
									22,4	0,92	0,00227
7	—	14	35	5	—	—	25 Co 5 Mo Rest Fe	700	8,2	0,13	0,00021
									12,4	0,68	0,00197
								650	15,3	0,64	0,00143
									18,0	1,23	0,00313
								600	20,3	0,3	0,0008
									22,	0,4	0,00047
8	—	15	60	—	2	—	7,18 Mo 15 Fe	600	15,2	1,25	0,0029
									19,8	3,4	0,0053
									25,3	Failure at 15%	—
							Anealed at 1300°		15,4	0,64	0,0026
									21,5	2,76	0,0093
								580	22,4	3,95	0,0026
9	0,58	14,9	19,5	3,8	1,8	3,6	3,8 Mo, Rest Fe	600	22,3	1,43	0,00427
								640	15,4	0,696	0,00207
									18,2	1,996	0,0061

Table 6.- Strength values of several austenitic steels at elevated temperatures.

No.	Composition in %						Tensile strength (kg/mm²) at temp. °C.				Alternating strength (kg/mm²) at temp. °C.			Creep strength (kg/mm²) at temperature °C.		
1[a]	C	Si	Mn	Cr	Ni	W	20°	400°	500°	600°	20°	500°	600°	400°	500°	600°
	0,56	1,58	0,52	15,5	13,3	2,02	95	75,5	72,9	66[b]	48,2	43,9	34,1[d]	69	36	17
							—	—	69,3	54[c]						
2[c]	0,07	0,65	0,5	18,25	9,65	—	20°	—	520°	640°	410°	520°	640°	300°	400°	500°
							60		42	33	± 23	± 23	± 21[f]	15	12	6
3[e]	0,125	0,58	0,47	18,5	9,57	—	64	—	47	34	± 26	± 27	± 23[f]			
4[g]	1,2	1,16	0,49	8,31	19,7	—	79	—	60	53	470°	600°	650°			
											± 32	± 26	± 25[f]			
5[h]	Fe	Mo					20°		400°	500°	400°	500°	600°			
	15	1	2	15	61	—	89	—	84	76	± 40	± 38	± 35			
6[h]	20	—	1	15	64	—	63	—	57	56	± 28	± 28	± 25			

a) M. Hempel and H. Timmermanns, Mitt. Kaiser-Wilhelm-Inst. Eisenforschung Dusseldorf, Report 308 (1936)
b) By short tearing test
c) After 20 minute test
d) Buckling strength
e) H. C. Cross, Trans. Amer. Soc. Mech. Engr. Vol. 46 (1934), S. 533/53
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Table 7.- Correlation of several AL-piston alloys for aircraft engines.

No.	State	Constituents in %								
		Cu	Si	Fe	Mn	Ni	Mg	Ti	Zn	Co
1	SG*)	4	—	0,6	—	2,0	1,5	—	—	—
2	S	2,25	—	1,4	0,5	1,3	1,6	0,1	—	—
3	G	10—11	—	0,45	—	1,0	0,25	—	0,12	—
4	GS	1,0	12	0,4	0,1	1,0	1,0	—	—	—
5	G	4,5	12	0,5—1,0	1,0	1,5	1,0	—	—	—
6	G	1,5	19	0,5	0,6	1,5	0,6	—	0,05	1

*) S = Forged, G = Cast

Table 8.- Effect of temperature on flexural fatigue strength of several alloys according to table 7 on the basis of 25.10^6 load reversals.

Alloy No.	State	Alternating strength in (kg/mm²) at temperature °C.		
1		20°	200°	300°
	G	5—6	5— 6	4,5
	S	14	13—14	9
3	G	7	6,5	5 —5,5
4	G	7— 8	7— 8	4,5— 5
	S	13—14	13—13,5	7,5
6	G	7	7	5

Table 9.- Effect of heat treatment of 100 hours at 200° on the strength at elevated temperatures of alloy No. 1 and 4 of table 7

Alloy No.	State	Tensile strength (kg/mm²)	Elong. %	Con-strac-tion	Anneal-ing period
1	S	29	4—14	19—26	30 min
		19	18—22	38—48	100 h
4	S	25	3—4	8,5—9,5	30 min
		12	15,5—18,5	25	100 h

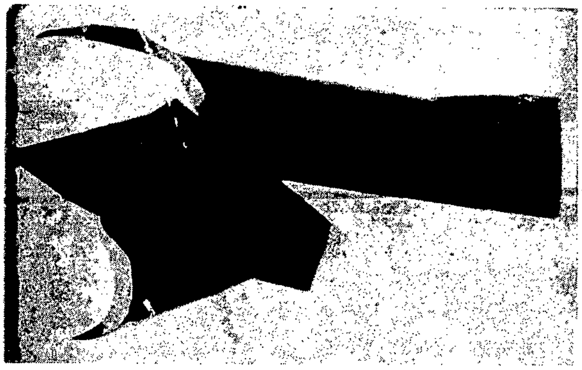


Figure 1.- Appearance of break on superheated steam engine blades.

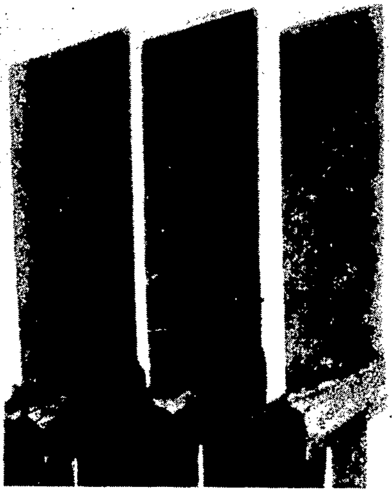


Figure 9.

Figure 10.

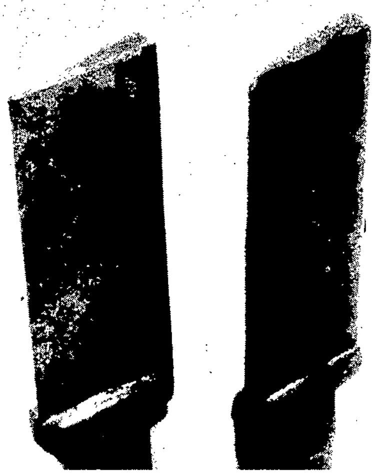


Figure 11.

Figures 9 to 11.- Deformation and fission due to repeated steep temperature changes.

Figure 9.-	Alloy No. 1,	270	temperature changes	Table 5
" 10.-	" No. 2,	212	" "	
" 11.-	" No. 8,	185	" "	

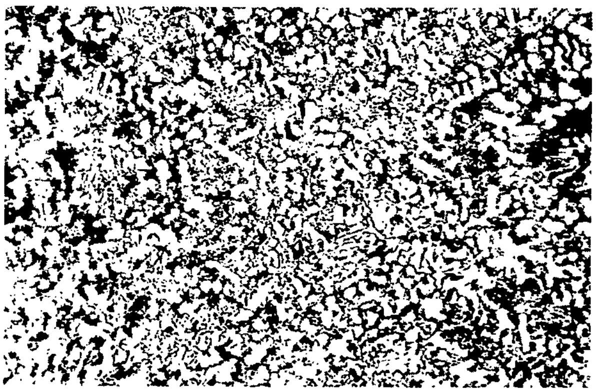


Figure 14.- Structure of the DVL-alloy, cast, × 100.

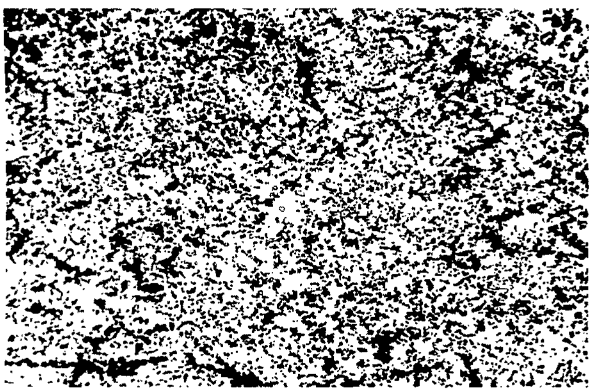
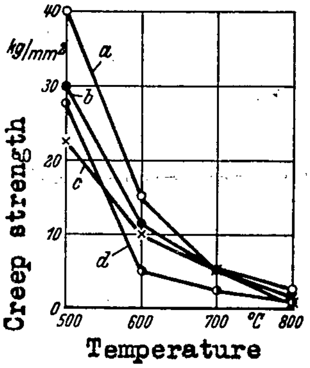


Figure 15.- Structure of the DVL-alloy, forged and precipitation hardened, × 100.



	C	Cr	Ni	W	Si
a	0,5	12	14	2	—
b	0,3	18	8	5	—
c	0,35	15	20	—	—
d	0,5	12	—	5	3

Figure 2.- Creep strength plotted against valve-steel temperature (1.10^{-3} percent/hour between 80th and 100th hour of load) according to Musatti and Regiori, Metallurgia Italiana, 1934, No. 7 to 10.

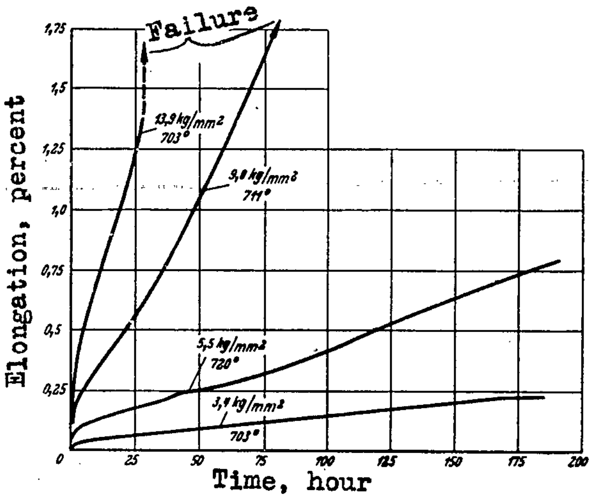


Figure 3.- Time yield curves for Co-W-Fe- alloys according to table 3.

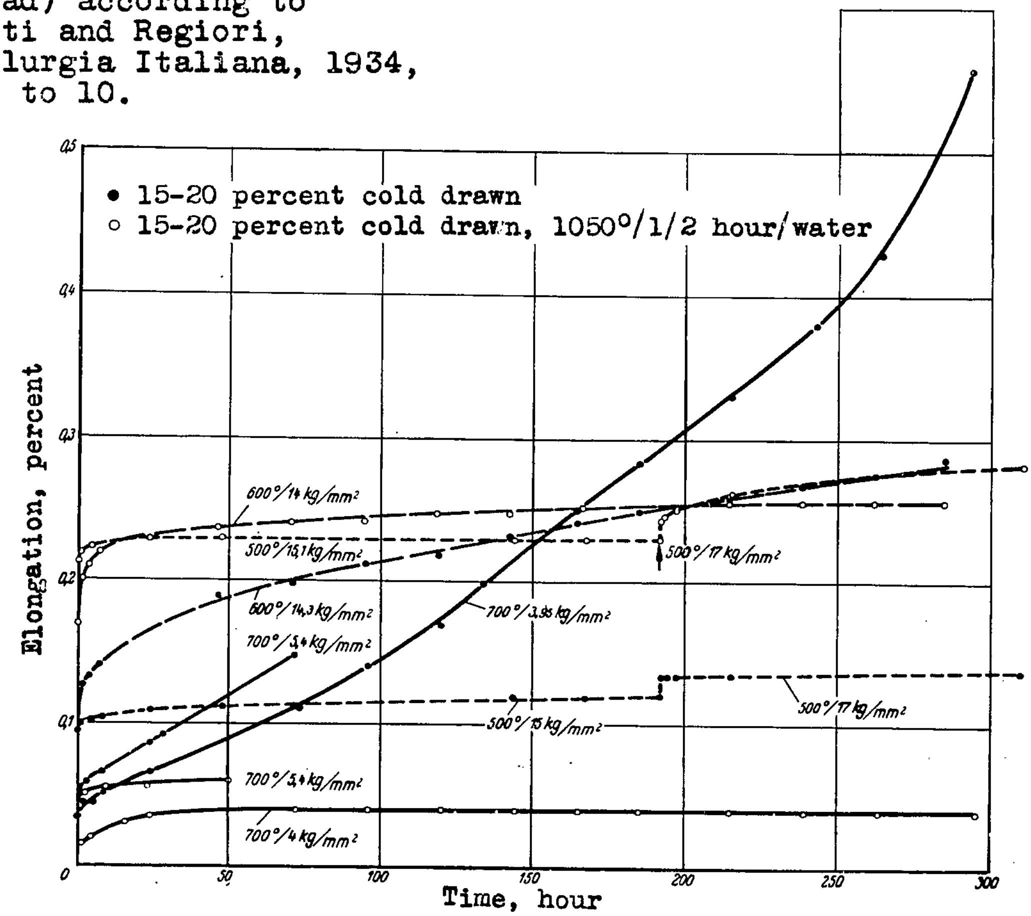


Figure 4.- Creep rate of steel No. 4 (table 4) in quenched and cold-worked state.

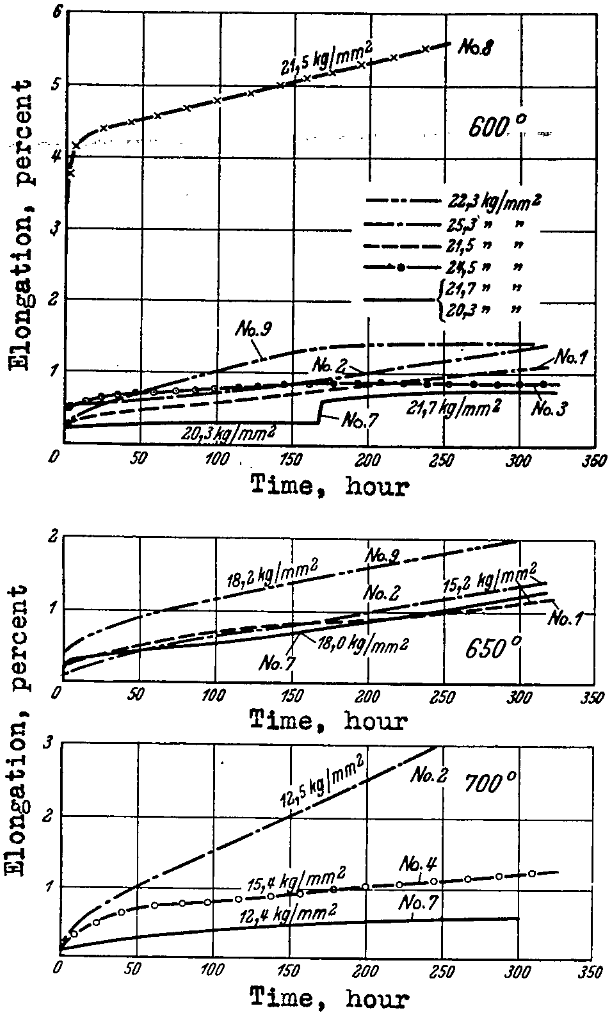


Figure 5.- Creep rate of several alloys of table 5 in creep test.

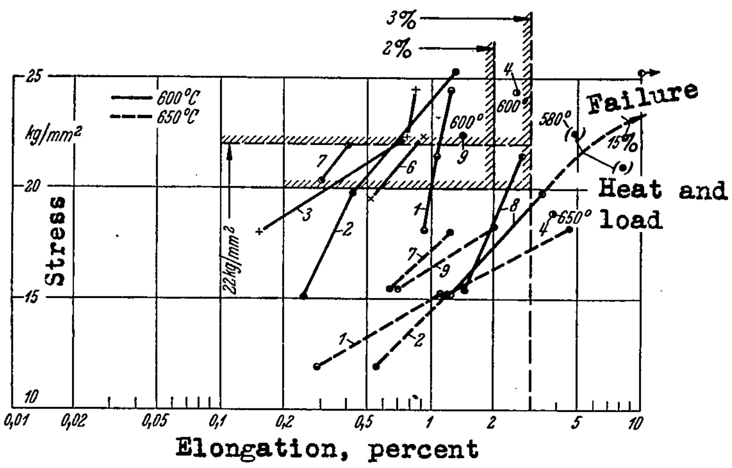


Figure 6.- Creep elongation of alloys of table 5 in 300 hour creep test.

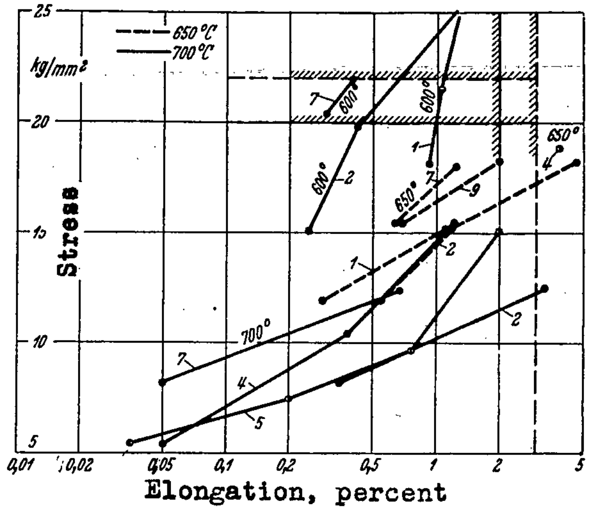


Figure 7.- Creep elongation of alloys of table 5 in 300 hour creep test.

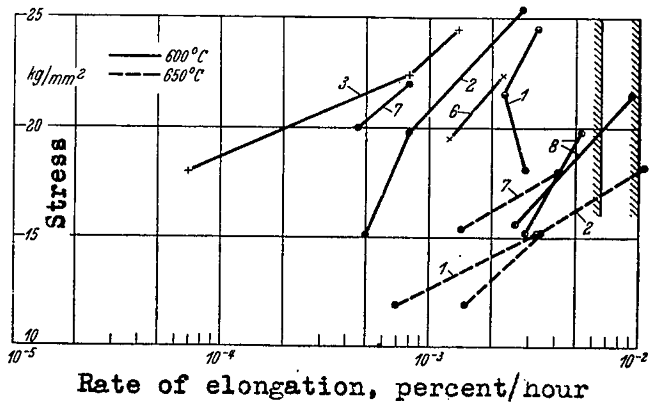


Figure 8.- Average creep rate for 300 hours of several alloys of table 6.

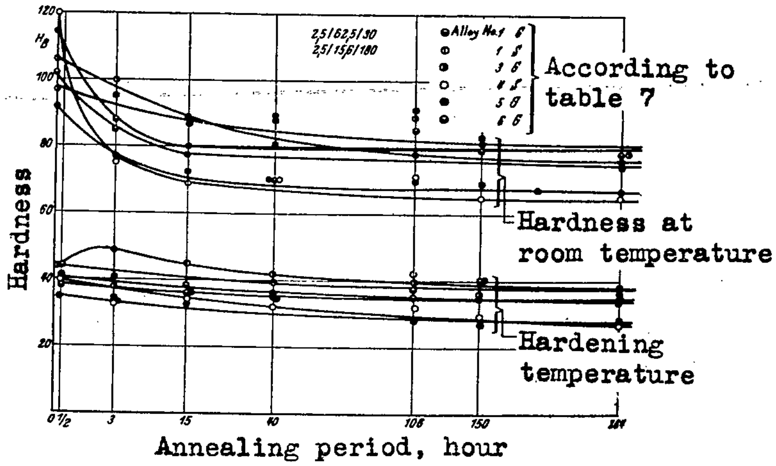


Figure 12.- Hardness at room temperature and hardening temperature after annealing at 250° C against annealing period for piston alloys.

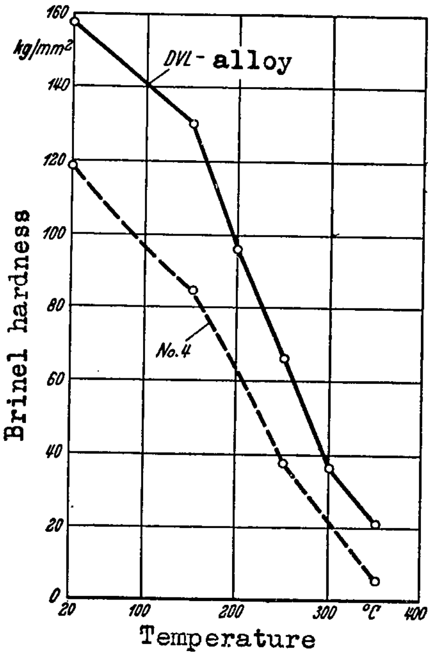


Figure 13.- Hardening temperature of the DVL-alloy at various temperatures.

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Author
Metals - Temperatures, High
Metals - Alloys - Temperatures, High
Metals - Alloys - Creep
Steels - Temperatures, High
Metals - Creep
Metals, Heat-resistant
Metals - Turbines, Gas
Metals - Engines
x Metals - Engine cylinder heads
Metals - Valves, Exhaust
Metals - Pistons
Steels - Creep